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Texture Development during Hot Deformation of Al/Mg Alloy Reinforced with Ceramic Particles

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Abstract

Al-3Mg alloy reinforced with B_4C particles in volume fractions of 5, 10 and 15% were subjected to hot deformation to investigate the impact of presence of ceramic particles and deformation condition on final texture. Single-heat hot compression test was performed at temperatures of 300 to 500 °C with strain rates of 10^{-3} to 10 s^{-1}. The electron backscatter diffraction method was applied to evaluate the final texture and microstructures. It was observed that the {110} fiber formed during deformation was intensified by increasing the Zenner-Holloman parameter, while deformation at lower *Z* makes {100} fiber pervasive throughout the matrix. Developing {100} fiber in such condition leads to continual softening of flow stress. Presence of particles by promoting particle stimulating nucleation mechanism at high *Z* and restricting grains rotation at low *Z* led to lower final texture intensity.

Key words: Al-Mg alloy, Texture, Hot deformation, Metal matrix composite, Texture softening

1. Introduction

Texture of microstructure is known as an important outcome of deformation at elevated temperature, which can affect the final product properties and performance such as strength, elongation [1], crack propagation [2,3] and fracture toughness [4,5]. Besides of material nature, it is well known that different processing parameters, i.e., heating rates during solution treatment [6], deformation strain, deformation temperature, deformation path [7] can also affect the evolution of texture. For example, in the case of aluminum, as a one of the most applicable metals in industry, it has been shown that adding Mg as the solute atom can change the texture by affecting the formation of shear bands [8] and intermediate annealing influences the final annealing texture of Al-Mg alloys [9]. Moreover, it was reported that the appearance of {110} and {001} components (compression plane) in Al-Mg alloys can be strongly depended on deformation temperature, and amount of strain and strain rate which controls dynamic restoration, grain growth, and deformation mechanisms [10].

Al-Mg alloys have been found different applications in product components which require moderate strength, formability, weldability, and corrosion resistance. It has been shown that although by increasing Mg, the ratio of strength to weight increases, the alloys containing more than 3% Mg due to formation of Al_3Mg_2 become sensitive to stress cracking corrosion. As the matter of fact presence of second phase into the microstructure in the form of ceramic particles or precipitate makes the texture evolution more complicated. In this way it was stated that more homogeneous deformation with higher texture intensity can be obtained by adding fine particles; in contrast, large particles lead to a more heterogeneous deformation and lower texture intensity [4]. It was also found that adding more particles although will not modify the main components of texture; it can reduce the intensity [11]. In addition to the effect of particle size and volume fraction on texture intensity, Shahani and Clyne [12] have shown that the texture intensity will be higher as the spherical particles are used to reinforce matrix than the elongated particles. The direct impact of particle-matrix interface strength on texture intensity has studied by Ramesh et al. [13]. Moreover, generally, it has been stated that coarse particles can promote recrystallization by particle stimulating nucleation (PSN) and fine particles by a Zener pining effect prevent grains recrystallization and even grain growth during deformation sequences [14,15].

The majority of previous researches have been focused on the effect of processing parameters on the texture evolution of single phase Al-Mg alloy, and seldom studies have been considered the impact of the second phase on final texture of Al-Mg alloy subjected to hot deformation. Hence, the aim of present work is to study the texture variation in Al-Mg/B_4C with different particles concentration for a wide range of deformation conditions.

2. Experiment

Al-3wt.% Mg alloy and Al-3wt.% Mg/B_4C composites with average particle size of 80 µm and volume fraction of 5, 10 and 15% were fabricated from stir-casting followed by hot extrusion which is described in details in previous work [16]. Chemical composition of Al-3Mg and composite, obtained via Electron Probe Micro-Analyzer (Cameca SX100) and Quantometer analysis, respectively, were around 2.9 Mg, 0.15 Fe, and 0.2 Si (wt.%).

For single-hit hot compression tests, cylindrical specimens with length of 11.5 and 10 mm and diameter of 7.4 and 5 mm were used for Instron 4507 universal deformation machine and Baehr DIL-805 deformation dilatometer, respectively. Deformation was performed at temperature of 300 to 500˚C with the constant strain rate of 10^{-3} to 10 s^{-1}.

In order to investigate the microstructure texture of the rapidly quenched samples (less than 2 seconds), electron backscatter diffraction (EBSD) method with Thermal-Field-Emission Scanning Electron Microscope (Jeol 7001f-0.1-30 kV) was employed. The orientation distribution function (ODF), calculated from EBSD results, are presented as plots of constant $\phi 2$ sections with isointensity contours in Euler space defined by the Euler angles $\phi 1$, φ, and $\phi 2$.

3. Result and discussion

3.1.Deformation condition and texture development

Fig.1a represents the orientation distribution functions (ODFs) of Al/Mg-10%B_4C composite in annealed condition for $\varphi 2=0°$ and $\varphi 2=45°$. According to this figure, the sample shows a cube texture with component of {100} <001> which is well known for annealed Al-Mg alloys. Based on Fig. 1b, after deformation at 400°C with strain rate of 0.1 s^{-1}, a strong fiber along $\phi=45$ for $\varphi 2=0$ is appeared. This fiber corresponds to volume elements with {110} plans parallel to the compression direction. It will be more sensible when the {110} and {111} figures are subjected to the investigation (Fig. 1c). It is clear that pole {110} is parallel to compression direction and pole {111} is 45° away from the deformation axis. It has been widely reported that for fcc metals subjected to uniaxial deformation, the high-density plane, i.e. {111}, will rotate to the direction with highest share stress, i.e. around 45°. This fact can be seen in {111} pole figure, Fig.1c. As a result, {110} plane will be along with normal stress direction, which here is compression direction. In other words, formation of fiber {110} is attributed to activation of {111}<110> slip system [10].

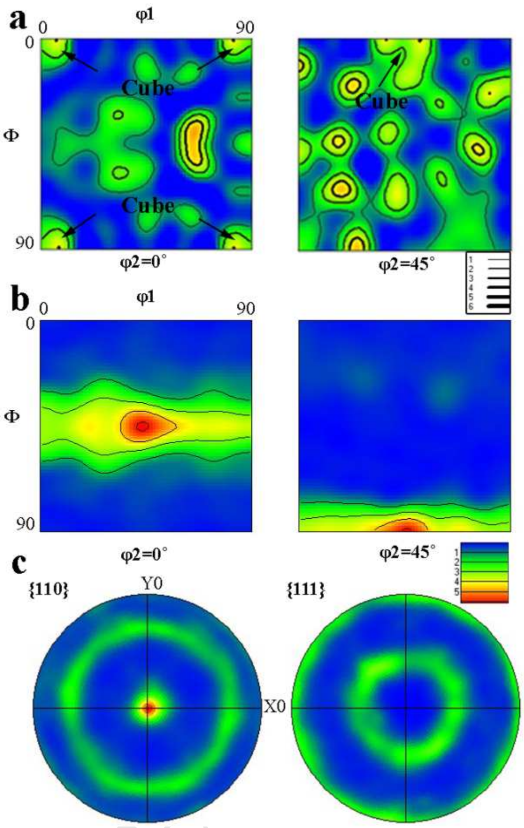

Fig.1: (a) Orientation distribution functions for annealed composite and (b, c) ODFs and pole figure of composite deformed up to strain 0.7 at 400 ˚C with strain rate of 10 s^{-1}.

Although deformation led to the appearance of a strong fiber along deformation axis, according to Fig. 2, it seems that the deformation condition, i.e., temperature and strain rate, has a great impact on the fiber orientation and intensity. According to this figure, as deformation temperature is raised or strain rate is decreased, the intensity of pole {101} is reduced. Moreover, it seems that the fiber orientation is rotating from <101> to <001>. It can be more visible if the average orientation densities, f(g), along the fibers of {110} and {100} were calculated and plotted against Zenner-Holloman parameter ($Z = \dot{\varepsilon}\exp(130000/RT)$), as the unified deformation condition parameter (Fig. 3). Based on these results, deformation at higher *Z*s intensifies {110} fiber while deformation at lower *Z*s makes {100} fiber pervasive throughout the matrix.

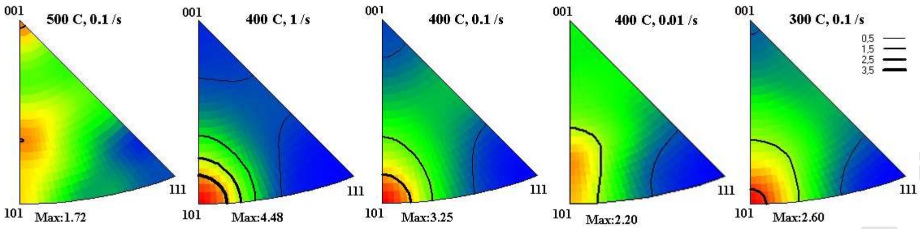

Fig. 2: Inverse pole figure representation for different deformation condition of Al-3Mg/10%B_4C composite.

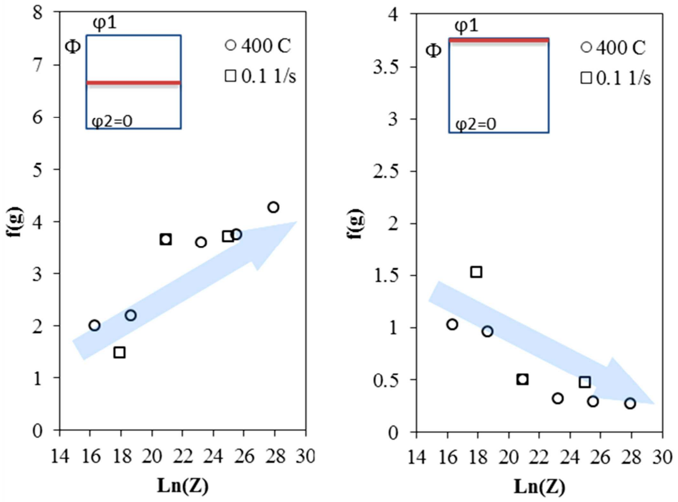

Fig. 3: variation of average orientation densities f(g) along the two fibers {110}, i.e. Φ=45 and {100} i.e. Φ=0 with respect to *Z*.

3.2.Texture and microstructure evolution

Fig.4 shows microstructure obtained after deformation at 500 ˚C with strain rate of 10^{-3} s^{-1}. Based on the fact that the initial grain size was around 30µm [17], the presence of large grains in the matrix are results of grain grow, i.e., grain boundary movement and grains rotation. As it can be seen in Fig.4, a large grain with orientation <001> parallel to compression axis has lower Schmid factor. According to the grain boundary curvature, it seems that this grain was growing toward other grains with higher Schmid factor. It has been reported that during deformation of Al-3Mg when the solute drag of dislocation motion is the dominate deformation mechanism, by increasing strain the fiber component changes from {110}+{001} to {001} and the fiber texture with component {001} is stable during deformation of Al-Mg [18]. Jeong et al. [10] suggested that the formation of {001} fiber is attributed to grain boundary migration of grains with {001} orientation. It is due to the higher recovery rate in these grains with respect to others. As the result, these grains have lower energy and they are stable. Moreover, these grains have lower Schmid factor which results in less deformation and lower dislocation density in the grains. Therefore, their grain boundaries move toward the other grains with higher energy. In other words, {001} grains grow during deformation at low strain rate and high temperature, i.e., low *Z*s ($<10^{-2}s^{-1}$), which leads to developing of {001} fiber.

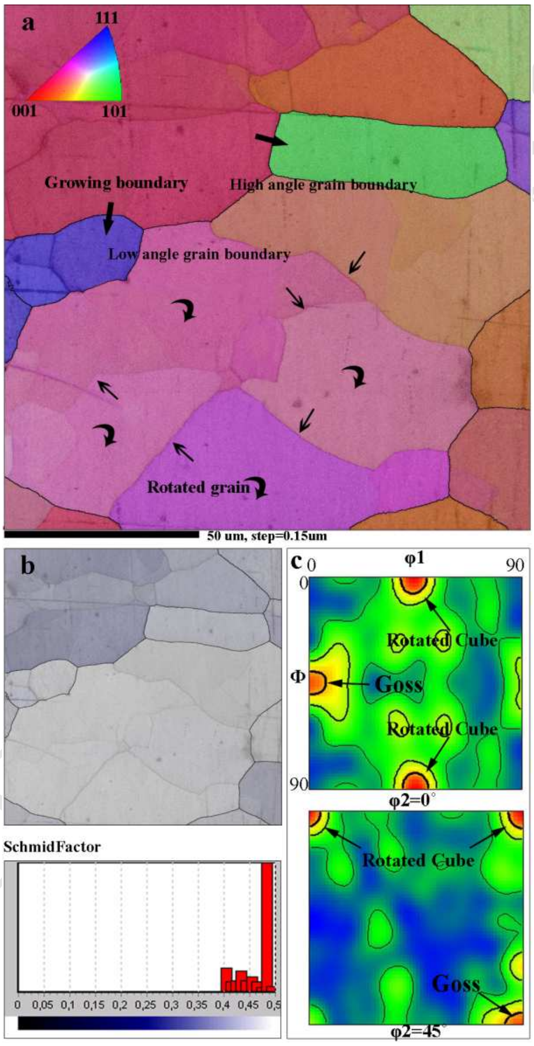

Fig. 4: a) EBSD micrograph, b) Schmid factor value in the corresponding grains and c) ODFs of composite deformed up to strain 0.7 at 500 ˚C with strain rate of 10^{-3} s^{-1}.

ODFs maps presented in Fig. 4c show that the deformed composite at low *Z* has rotated cube {001}<1-10> and Goss {011}<100> textures, while composite deformed at higher *Z* had

{011} fiber texture (Fig.1). Based on what was mentioned so far, at the low value of *Zs*, the appearance of cube component was expectable, and the crystal rotation is because of shear deformation as the results of the presence of friction [7]. Moreover, according to Ramesh et al. [13] the appearance of Goss component in Al matrix composite can be attributed to DRX during deformation.

Considering to the previous work [17], in some deformation situations (middle *Zs*), particle stimulating nucleation (PSN) mechanism led to the appearance of fine DRXed grains in the vicinity of B_4C particles (Fig. 4). As it can be seen from Fig. 5, based on the volume fraction of DRXed grains, it is not expected that the DRXed grains have a significant effect on texture intensity. But it should be noted that increasing particles has three aspects, 1) enhancing deformation zones area and appearance of high energy area as a results of overlapping deformation zones which lead to more recrystallization, 2) reducing volume fraction of matrix with deformation texture, and 3) increase the volume fraction of distorted area around particle in case of none-DRX condition. In fact, even if PSN was not active during deformation as matrix flow around particles during deformation, particles constrain led to change the deformation path and hence deformation texture. To evaluate the impact of DRXed grains on the dominate texture, the inverse pole figures of Al-3Mg single phase alloy and Al-3Mg matrix composites reinforced with two different volume fractions of B_4C, deformed in the same condition are compared in Fig. 6. According to this figure although in three cases {101} fiber is dominated, the intensity of texture was reduced by applying more reinforcement. In other words, the randomly oriented grains which are results of dynamic recrystallization, weaken the texture.

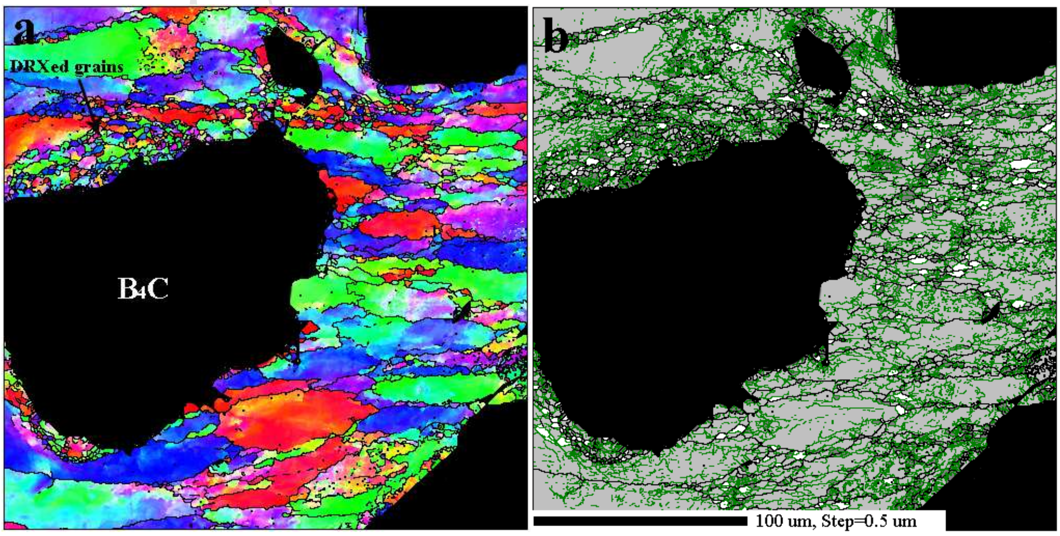

Fig.5: DRXed grains around particles as the results of PSN mechanism for Al-3Mg/10%B_4C deformed at 400 ˚C with strain rate of 0.1 s^{-1}.

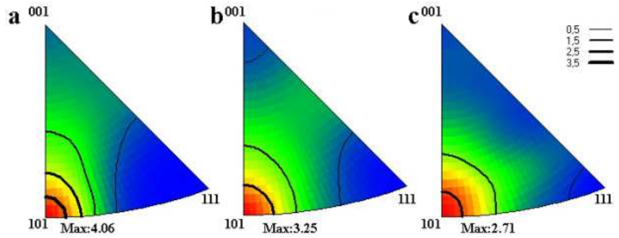

Fig.6: Inverse pole figures for a) single-phase alloy, b) Al-3Mg/10%B_4C and c) Al-3Mg/15%B_4C composites deformed at 400 ˚C with strain rate of 0.1 s^{-1}.

3.3. Texture development and flow stress

According to Fig.7, despite the change in reinforcement volume fraction, the flow stress decreases by straining, which is typical during dynamic recrystallization. Similar results have been reported in previous works [19-22] which mainly related to the occurrence of DRX, rearrangement of particles and impurities along flow lines and transformation of dislocation wall structure to subgrain boundaries. Based on microstructure investigations presented in previous work [17] in such deformation situation, i.e., high temperature and low strain rate, dynamic recrystallization will not be dominant, and so it cannot affect the flow stress significantly. Moreover, based on the fact that composites with the different volume fraction of reinforcement behave similar to single-phase alloy, the idea of particles rearrangement softening cannot be correct. However, grains and subgrains growth was observed during deformation at very low *Z*s which it, in turn, can reduce the flow stress.

The other aspect of grain growth is texture variation. As it was explained, by deformation at low *Z*s, the {101} fiber component changed to {001}. It has been stated that the single crystal of aluminum with component {001} parallel to compression direction has lower strength than {011} single crystal [23]. As it was mentioned in the previous section, grain growth is responsible for changing {101} fiber to {001} fiber which needs strain. Therefore, as the development of texture with strain is gradual, continual softening of flow stress seems reasonable. For composite containing particles with appropriate matrix-particle interface strength, the grains adjacent to particles will not rotate and growth easily. Grains rotation (as shown in Fig. 4) is due to increase in grain's Schmide factor which as mentioned before, improves the {001} grains growth rate by providing high dislocation density grains. Therefore, the overall {001} fiber intensity will be lower than the single phase alloy and hence the softening rate will decrease. Fig. 8 represents the Schmid factor variation in the microstructure of different samples. It is obvious that grains inside the matrix and far away from particles have higher Schmid factor which is due to their ability to accommodate with deformation axis, however, grains in the vicinity of particle have lower Schmid factor, especially those stuck to the particles.

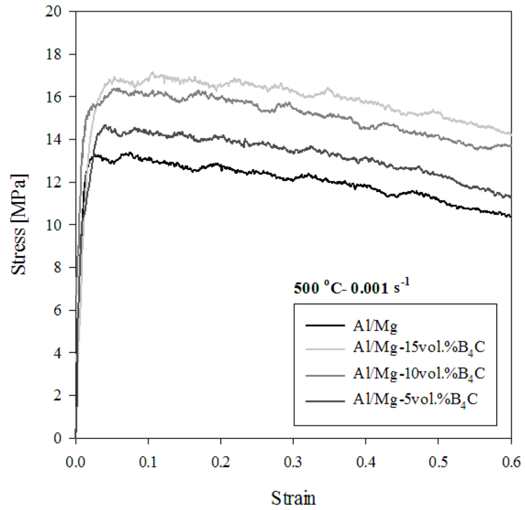

Fig. 7: True strain-stress curves obtain from hot compression test at 500 ˚C with strain rate of $0.001s^{-1}$.

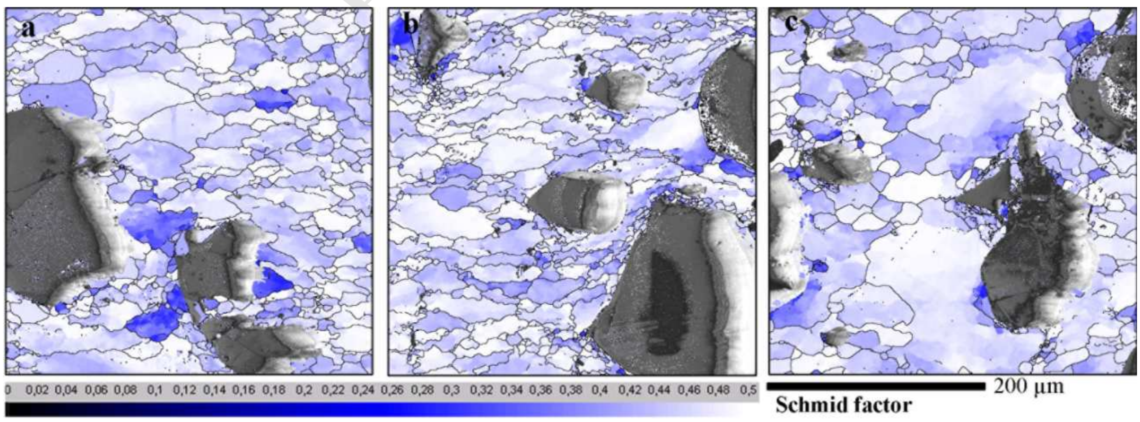

Fig. 8: The variation of Schmid factor through grain in samples deformed at 400 ˚C and a) with strain rate of $0.001s^{-1}$ and b) $0.\ 01s^{-1}$ and c) at 500 ˚C with strain rate of $0.1s^{-1}$.

Conclusion

Al-3Mg alloy reinforced with B_4C particles in volume fractions of 5, 10 and 15% were subjected to hot deformation to investigate the impact of second phase and deformation condition on final texture.

1) Formation of {110} fiber during deformation at high value of Zenner-Holloman parameter is attributed to activation of {111}<110> slip system.
2) The growth of grains with {001} component which has lower Schmid factor leads to developing {001} fiber during deformation at low strain rate and high temperature.
3) Presence of particles by promoting particle stimulating nucleation mechanism at high *Z* and restricting grains rotation and growth at low *Z*, lowers final texture intensity.
4) Developing {100} fiber by straining at low *Z*, cause a continually softening of flow stress.

References

[1] Z.J. Wang, M. Ma, Z.X. Qiu, J.X. Zhang, and W.C. Liu, "Microstructure , texture and mechanical properties of AA 1060 aluminum alloy processed by cryogenic accumulative roll bonding," Materials Characterization, vol. 139, 2018, pp. 269-278.

[2] F. Li, Z. Liu, W. Wu, P. Xia, P. Ying, and Q. Zhao, "On the role of texture in governing fatigue crack propagation behavior of 2524 aluminum alloy," Materials Science & Engineering A, vol. 669, 2016, pp. 367-378.

[3] W. Wu, Z. Liu, Y. Hu, F. Li, and S. Bai, "Goss texture intensity effect on fatigue crack propagation resistance in an Al-Cu-Mg alloy," Journal of Alloys and Compounds, vol. 730, 2018, pp. 318-326.

[4] S. Chatterjee, S. Koley, S.D. Bakshi, and M. Shome, "Role of crystallographic texture , delamination and constraint on anisotropy in fracture toughness of API X70 line pipe steels," Materials Science & Engineering A, vol. 708, 2017, pp. 254-266.

[5] Q. Zhao, Z. Liu, T. Huang, P. Xia, and F. Li, "Enhanced fracture toughness in an annealed Al-Cu-Mg alloy by increasing Goss / Brass texture ratio," Materials Characterization, vol. 119, 2016, pp. 47-54.

[6] X. Wang, M. Guo, L. Cao, J. Luo, J. Zhang, L. Zhuang, "Effect of heating rate on mechanical property , microstructure and texture evolution of Al – Mg – Si – Cu alloy during solution treatment," Materials Science & Engineering A, vol. 621, 2015, pp. 8-17.

[7] F.J. Humphreys and M. Hatherly, “Recrystallization and related annealing phenomena,” Elsevior, 2nd, 2004.

[8] M. Koizumi, “Role of Shear Band in Texture Control of AI-Mg Alloys,” Metals and Materials vol. 5, 1999, pp. 1-7.

[9] M.A. Wells, D.J. Lloyd, I.V. Samarasekera, J.K. Brimacombe, and E.B. Hawbolt, “Modeling the microstructural changes during hot tandem rolling of AA5XXX aluminum alloys : Part II . Textural evolution,” Metallurgical and Materials Transactions B, vol. 29, 1998, pp. 621-633.

[10] H.M. Jeong, K. Okayasu, and H. Fukutomi, “{001} Texture Map of AA5182 Aluminum Alloy for High Temperature Uniaxial Compression,” Materials Transactions, vol. 51, 2010, pp. 2162-2167.

[11] A. Poudens, B. Bacroix and T. Bretheau, “Influence of microstructures and particle concentrations on the development of extrusion textures in metal matrix composites,” Materials Science & Engineering A, vol. 196, 1995. Pp. 219-228.

[12] R.A. Shahani and T.W. Ciyne, “Recrystallization in fibrous and particulate metal matrix composites,” Materials Science and Engineering A, vol. 135 , 1991, pp. 281-285.

[13] C.S. Ramesh, R. Keshavamurthy, P.G. Koppad, and K.T. Kashyap, “Role of particle stimulated nucleation in recrystallization of hot extruded Al 6061/SiCp composites,” Transactions of Nonferrous Metals Society of China, vol. 23, 2013, pp. 53-58.

[14] F.J. Humphreys, “The thermomechanical processing of A1-SiC particulate composites,” Materials Science and Engineering A, vol. 135, 1991, pp. 267-273.

[15] F.J. Humpherys and P.N. Kalu, “Dislocation-particle interactions during high temperature deformation of two-phase aluminium alloys,” Acta Metallurgica, vol. 35, 1987, pp. 2815-2829.

[16] M. Rezayat, M.R. Bahremand, M.H. Parsa, H. Mirzadeh, and J.M. Cabrera, “Modification of As-cast Al-Mg / B4C composite by addition of Zr,” Journal of Alloys and Compounds, vol. 685, 2016, pp. 70-77.

[17] M. Rezayat, M.H. Parsa, H. Mirzadeh, and J.M. Cabrera, “Microstructural investigation of Al-Mg /B4C composite deformed at elevated temperature,” Journal of Alloys and Compounds, vol. 763, 2018, pp. 643-651.

[18] K. Okayasu, H. Takekoshi, and H. Fukutomi, “Influence of grain boundary migration on { 001 } texture formation in Al-3 mass % Mg based alloys during high temperature compression deformation,” Materials Transactions, vol. 48, 2007, pp. 2002-2007.

[19] J.R. Cho, W.B. Bae, W.J. Hwang, and P. Hartley, “A study on the hot-deformation behavior and dynamic recrystallization of Al ± 5 wt .% Mg alloy,” Journal of Materials Processing Technology, vol. 118, 2001, pp. 356-361.

[20] K. Ihara and Y. Miura, “Dynamic recrystallization in Al–Mg–Sc alloys,” Materials Science and Engineering, vol. 389, 2004, pp. 647-650.

[21] H.J. Mcqueen and I. Poschmann, “Subgrain development in hot working of Al and Al-5Mg,” Materials Science & Engineering A, vol. 234, 1997, pp. 830-833.

[22] I. Poschmann and H.J. Mcqueen, “Flow softening and microstructural evolution of al-5mg during hot working,” Scripta Materialia, vol. 35, 1996, pp. 1123-1128.

[23] S. Gourdet and F. Montheillet, “An experimental study of the recrystallization mechanism during hot deformation of aluminium,” Materials Science and Engineering A, vol. 283, 2000, pp. 274 - 288.

Research Highlights

1) In Al-Mg developing {100} fiber at low *Z*, leads to continually flow softening
2) {111} and {100} fibers develop via; dislocation sliding and preferred growth
3) Lower texture intensity for various Z is the output of adding particles to the matrix
4) Presence of particles in matrix promotes PSN at high *Z* and restricts GBS at low *Z*